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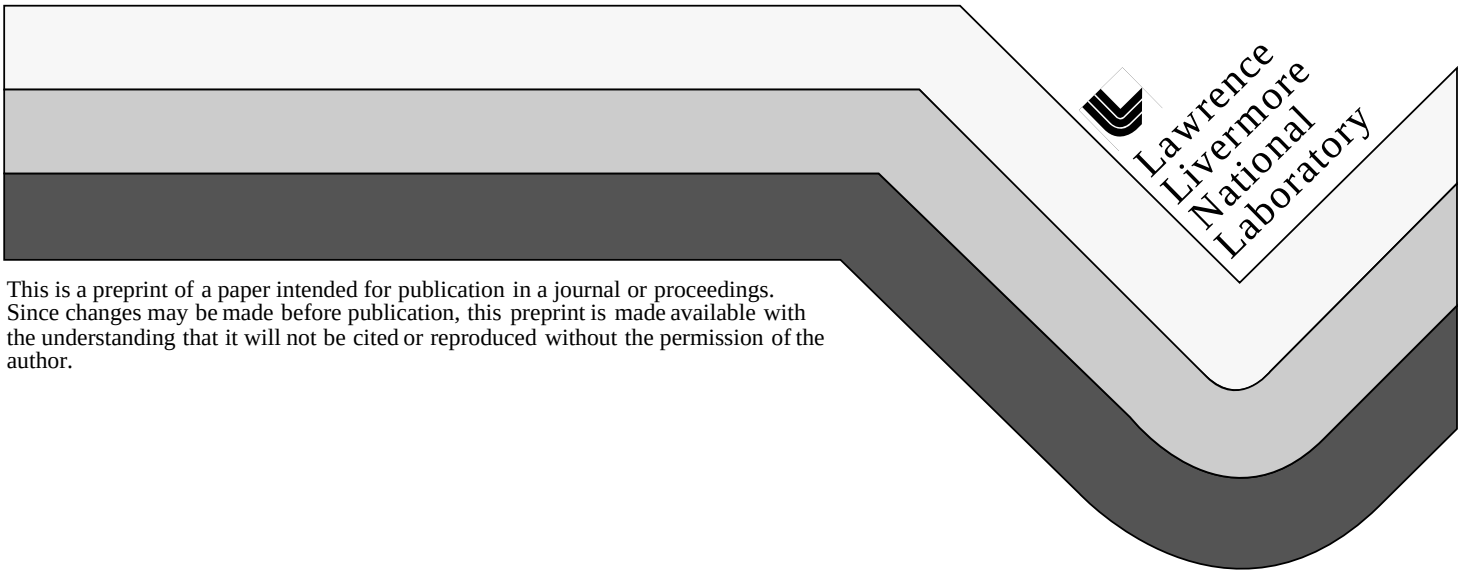

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INTERFACE CONTROLLED DEFORMATION TWINNING IN AN INTERMETALLIC NANOLAMINATE

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Microstructure

Figure 1 is a bright-field TEM image showing a typical edge-on microstructure within a TiAl (γ)-Ti_3Al (α_2) intermetallic nanolaminate. In general, the material contains two types of interfaces [1]: (1) The γ/α_2 interphase interface which has a usual orientation relationship $(0001)_{\alpha 2} \parallel (111)_\gamma$ and $<11\bar{2}0>_{\alpha 2} \parallel <1\bar{1}0>_\gamma$. (2) The γ/γ twin-related interface which includes true-twin (180° rotational) and pseudo–twin (60 and/or 120° rotational) interfaces. Here, the width of α_2 layers ranges from 10 to 50 nm, and that of γ layers ranges from 150 to 300 nm. Figure 2 is a weak-beam dark-field (WBDF) TEM image showing a typical dislocation substructure within the nanolaminate. Both lattice dislocations (*LD* hereafter) within γ layer and a high density of interfacial dislocations (*ID* hereafter) on inclined interfaces can be clearly seen. The density of *ID* is much greater than that of *LD*, and the *LD* are primarily threading dislocations which terminate their two ends at the interfaces. While the *ID* on semi-coherent γ/α_2 and γ/γ pseudo-twin interfaces are *1/6*<112> or *1/3*<112> type misfit dislocations [2], the on γ/γ true-twin interface are mainly *1/6*$[11\bar{2}]$ type twinning dislocations or geometry necessary dislocations for accommodating the departure of true-twin interface from the exact (111) twin plane.

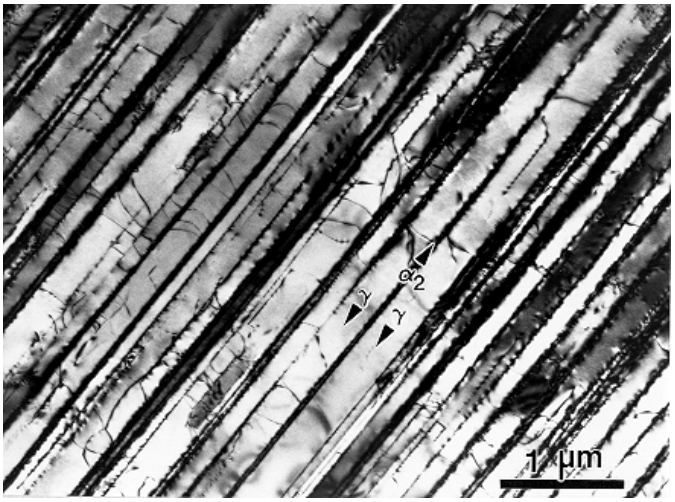


Fig. 1. Bright-field TEM image showing a lamellar grain viewing from an edge-on orientation.

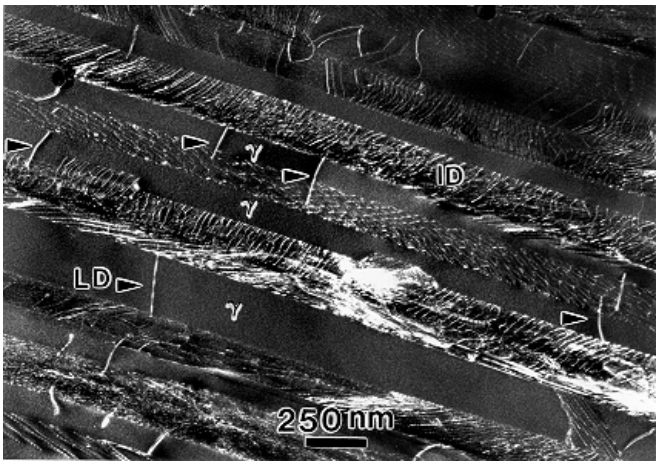
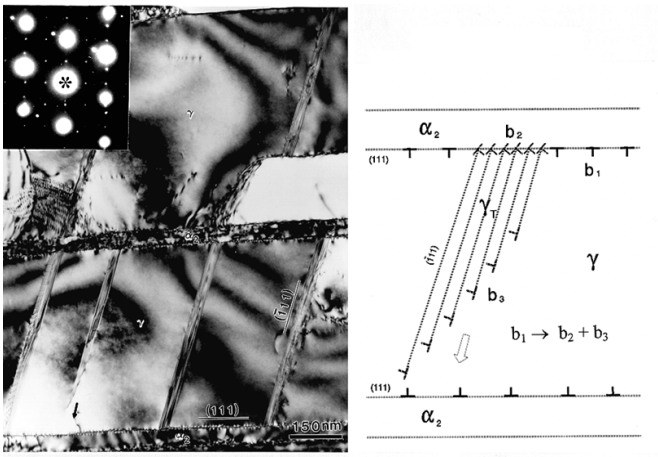


Fig. 2. WBDF image showing a typical dislocation structure of an inclined nanolaminate.

Deformation twinning and proposed mechanisms

When the nanolaminate was creep deformed at 760° C and 518 MPa, a deformation substructure associated with deformation twins (*DT* hereafter) within γ layers was developed. A typical example of the formation of $(\bar{1}11)$ [211]-type *DT* within the nanolaminate is shown in Fig. 3 (a). It is noted that one of the twin lamellae was still growing between two lamellar interfaces, and its growth would be eventually terminated by the lower interface. This observation suggests that the interfaces are preferred nucleation sites for *DT*, presumably resulting from the high local stresses caused by the pileup of *ID*. Accordingly, it is proposed that deformation twinning in the TiAl-Ti_3Al nanolaminate can be viewed as a stress relaxation process to relief the local stress concentration caused by the pile-up of interfacial dislocations during deformation. The effective stress (τ_e) at the tip of the pile-up of n dislocations can be evaluated by $\tau_e = n\tau_i$ [3], where τ_i is the resolved shear stress acting on the interface. To relieve the stress

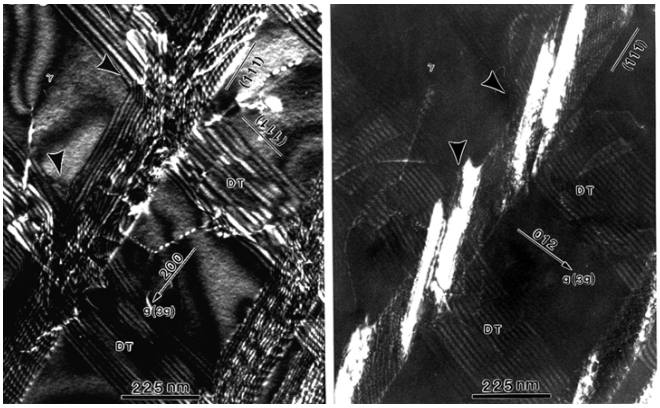
concentration, deformation twinning in γ layers is therefore taking place by a dislocation reaction based upon a stair-rod cross-slip mechanism [4,5]. As for an example of the $(\bar{1}11)$-type *DT* formed in the nanolaminate, the corresponding dislocation reaction (dissociation) is proposed to be $1/6[\bar{1}2\bar{1}]_{(111)} \rightarrow 1/6[011]_{(100)} + 1/6[\bar{1}1\bar{2}]_{(\bar{1}11)}$. The $(\bar{1}11)$-type *DT* is accordingly formed by a successive cross-slip of the twinning dislocations $1/6[1\bar{1}2]$ on the $(\bar{1}11)$ plane and leaving the stair-rod dislocations $1/6[011]$ on the (100) plane. Twin (stacking) faults are subsequently formed on the interfaces between the γ layer and *DT*. This is schematically illustrated in Fig. 3 (b).



(a) (b)

Fig. 3. (a) A bright-field TEM image showing several $(\bar{1}11)$ type deformation twins formed growing process toward another interface. (b) Schematic illustration of the nucleation of a $(\bar{1}11)$ type *DT* from a γ/α_2 interface, where b_1, b_2, and b_3 denote the interfacial, stair-rod, and twinning dislocations, respectively.

The formation of stair-rod dislocations at the intersections between the *DT* and α_2 layer is evidenced in Fig. 4, where the array of $1/6[011]$ stair-rod dislocations become invisible [Fig. 4(a)] or visible [Fig. 4(b)] when a reflection vector (***g***) *200* or *021* is used for imaging. It is noted that the individual stair-rod dislocation is not resolvable because of a narrow distance (0.25 nm) between two stair-rod dislocations. The significance of the proposed mechanism is to reveal that there are several barriers to be overcome in order to activate the twinning reaction. These barriers include (1) the repulsive force (F) between the interfacial (Shockley) and stair-rod dislocations, (2) the increase of line energy due to the dislocation dissociation, and (3) the increase of interfacial energy due to the formation of twin faults. Among them the repulsive force (F) between the interfacial (Shockley) and stair-rod dislocations is considered to be rate controlling. That is, a critical (minimum) stress (τ_C) is required to activate the dissociation reaction for twinning.



(a) (b)

Fig. 4. Paired WBDF images demonstrating the existence of the array of $1/6[011]$ stair-rod dislocations at the intersections (indicated by arrows) between the $(\bar{1}11)$-type *DT* and α_2 layer. (a) Invisible at $\boldsymbol{g} = 200$ ($\boldsymbol{g} \cdot \boldsymbol{b} = 0$), (b) visible at $\boldsymbol{g} = 021$, **Z** (zone axis) $\approx [0\bar{1}2]$.

Acknowledgments

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